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This program is being conducted for the purpose of applying the principle of rapid solidification to aluminum and iron alloy powder and subsequent development of stronger alloy compositions for fan blade application (Al alloys) and higher speed bearing materials (Fe alloys). Centrifugal atomization and forced convective cooling are being used to produce the fast cooled powder. During this report period, adaptation of the RSR process to aluminum and iron systems was continued. The evaluation of the first aluminum alloy matrix was partially completed, and a second matrix was selected. Hardness testing and microstructural evaluation of Fe alloys was performed.

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SECTION I

INTRODUCTION

Rapid solidification of metal alloys has shown that distinct solidification microstructures* can be attained with properties exceeding those available from conventional methods of solidification. These structures are recognized as a means of achieving major improvements in metal strength, alloy environmental compatibility, and fracture properties. Through the use of fast cooling, the following appear to be possible: (1) tougher and more corrosion resistant steels because of finer grain size and improved homogeneity, and (2) a new generation of aluminum, copper, and nickel alloys with decreased microsegregation and improved secondary phase dispersion.

An Advanced Research Project Agency (ARPA) sponsored program with the Pratt & Whitney Aircraft Group, Government Products Division (P&WA/Florida), has shown that the P&WA RSR process and equipment will achieve rapid solidification in spherical powder under steady-state operating conditions commensurate with production rates in excess of 1400 lb/hr. Further, this program has demonstrated concurrently, that high product quality can be achieved and that the resulting powder metal is in a form which can be readily handled and processed into useful shapes for subsequent application.

The program has also demonstrated that controlled, rapid solidification can produce a microcrystalline alloy structure with less microsegregation than RSR dendritic structures. It has also shown that a central rotary source can be used for liquid metal atomization into powder particles of sizes commensurate with average particle cooling rates of 10^5-$10^{6\circ}$K/second.

This program is a modification to the ARPA-sponsored work which is directed toward superalloy development. Its purpose is to expand the scope of work in the field of rapid solidification from the exclusive study of superalloys to a study of aluminum- and iron-base alloys. The specific objectives of this added effort are the development of an improved aluminum alloy suitable for V/STOL-A fan blades and an improved iron alloy suitable for rolling element bearings for advanced aircraft powerplants.

The program is a 36-month effort which begins with adaptation of the rapid solidification rate process to aluminum and iron alloy systems and terminates with a payoff analysis of new materials adapted to V/STOL-A and F100 advanced engine derivative requirements. This is the fourth technical report and covers the tenth through the twelfth months of the program. It deals with adaptation of RSR processing to aluminum and iron systems and the subsequent evaluation of these alloys.

*Kear, B. H., P. R. Holiday and A. R. Cox, Metall. Trans. 10A (1976) 191.

SECTION II

MATERIALS EVALUATION

Two alloy matrices are currently being evaluated and are listed in Table 1.

TABLE 1. SERIES I ALUMINUM AND IRON MATRIX

Al Matrix*

	Cobalt			
Zn	0.8	1.6	2.4	3.2
5.6	X	X		X
7.0	X	--	--	X
8.4	X	X	--	X
9.8	X		X	--

Fe Matrix*

		Molybdenum			
Cr	C	2	4	6	8
4	0.8	X	X	--	X
			(M-50)		
9	0.95	X	---		X
14	1.1		X	X	--
		(EX-00007)			
19	1.25	X	--	---	X

Fe, V held constant

*Amounts are in wt %.

In the aluminum matrix, Zn and Co were varied using similar or greater amounts of Zn than specified in 7075 and concentrations of Co from 0.8 to 3.2 wt %. Cobalt forms an interdendritic Co_2Al_9 phase which should effectively control grain size during elevated temperature deformation producing enhanced mechanical properties. Greater amounts of Zn were added to increase the concentration of age hardening precipitates for higher yield strength. During this reporting period, the evaluation of the first aluminum alloy matrix was completed, but selected compositions were repeated to determine if the Co_2Al_9 size and distribution can be produced on an extremely fine scale. Process refinements leading to improved dispersions of other thermally stable phases will be further developed during the course of this program.

One of the overall program goals is to develop an aluminum fan blade alloy that will economically replace existing titanium alloys. The 10 to 11 million psi elastic modulus inherent to cobalt-modified 7075 alloys, along with anticipated poor elevated temperature properties, make this alloy system unattractive for a fan blade application. First bending mode vibrational stresses cannot be dampened effectively in a low modulus material without an excessive number of airfoil shrouds. Thus, any new candidate aluminum alloy will need a specific modulus equivalent to or better than titanium alloys along with a comparable elevated temperature 450°F (232°C) strength-to-weight ratio. These requirements could be satisfied by dispersion strengthening with high volume fractions of fine, thermally stable intermetallic phases having high elastic modulii. Recent work performed by Alcoa* and Rostoker et al.,† has shown that transition metal additions to aluminum alloys are effective in increasing specific modulus and tensile properties up to 450°F (232°C). It is believed that the full potential of these promising alloy systems will be realized by fully inert RSR powder processing. As a result, a second aluminum alloy series has been formulated which will come closer to meeting the overall program goal.

* Sanders, R. E., Jr. et al., "Elevated Temperature Al Alloy Development," Contract No. F33615-77-C-5086, 31 March 1979.

† Rostoker, W. et al., "Fast Freezing As a Method for Al Alloy Development," Contract No. F33615-70-C-1525, 31 December 1972.

These alloys will investigate the use of two different strengthening strategies:

1. Binary peritectic alloy systems capable of forming strongly supersaturated solutions which may produce elevated temperature precipitation.

2. Ternary and quaternary alloy systems which form coupled eutectic structures in which the aluminum matrix phase can sustain supersaturation with consequent elevated temperature precipitation hardening.

A successful elevated temperature alloy should demonstrate a low rate of structural coarsening. Two factors contribute to this low coarsening rate. First, intermetallics formed by either decomposition of a supersaturated solid solution, or from a eutectic reaction, should have a low interfacial energy. Knowledge of specific eutectic systems and data from the relationship between interparticle spacing and growth rate allow interfacial energies to be calculated. Second, mass transport rates in solid aluminum should be low, implying the need to use diffusing species with high diffusion activation energies and hence, lower diffusivity. Work on aluminum alloys[*] has identified aluminum intermetallics with Mo, Fe-W, and Fe-Mo as potentially attractive intermetallic phases. These phases have a cubic WAl_{12} structure and therefore occur in higher volume fractions than intermetallics based on similar cobalt contents with the Co_2Al_9 structure. The diffusion rates of Fe, Mo and W in aluminum are slower than Co, so that coarsening of intermetallics based on Fe, Mo and W during fabrication should be slow.

The inclusion of iron in concentration of 6 to 8% produces eutectic $Al-Al_6Fe$ in rapidly solidified powders, and the possibility of further precipitation hardening the alloy with (Fe, Mo) Al_{12} precipitates from the supersaturated matrix is currently under investigation. Alloy development with Fe, either in Al_6Fe intermetallics, or by Fe-Mo in Al_{12} (Fe-Mo) is thus theoretically feasible.

Conventional precipitation hardening of the aluminum matrix of a fibrous eutectic microstructure is feasible[†], and suitable compositions for $Al-Al_6Fe-MgZn_2$ precipitation hardening systems have been investigated, showing tensile strengths above 110 ksi. A range of different precipitation species will be investigated, and the compositions outlined in Table 2 outline further binary alloy development toward meeting this program goal.

TABLE 2. SERIES II ALUMINUM ALLOYS[*]

High Liquidus Temperature Binary Alloys

Al – 5% W
Al – 5% V
Al – 5% Mo

Ternary and Quaternary Matrix Precipitation Hardening Alloys

Precipitation Elements	Iron		
	3	5	7
2.5 Mg-7Zn-2Cu	X		X
2 Mo		X	X
2 Si			X
5 Si		X	
16 Si-2 Mg	X		

*Amounts are in wt %.

* Mondolfo, L. F., "Aluminum Alloys, Structure and Properties," Butterworths, London 1976.
† Keong, P. G., J. A. Sames, C. McL. Adam and R. M. Sharp, "Solidification and Casting of Metals," p. 110, The Metals Society, London 1979.

SECTION III

CONVERSION AND CONSOLIDATION

Nine aluminum alloy powder runs were made during this report period to complete the matrix of Zn and Co additions to 7075 alloy. These conversions were made with the compositions listed in Table 3, and with data pertinent to the operations contained in Table 4. Typical yields of −140 mesh powder were similar to the yield reported during the last quarter of 50% or better. Run XSR 89 was performed using an atomizer speed of 35,000 rpm instead of the standard 24,000 rpm. This resulted in a −140 mesh powder yield of 81%, the highest obtained to date.

TABLE 3. COMPOSITIONS OF ALUMINUM ALLOYS CONVERTED TO POWDER (wt %)

VM No	Zn	Mg	Cu	Co	Al	XSR Run No.
716	5.6	2.5	1.0	1.6	Bal	78, 79, 80
717	5.6	2.5	1.0	3.2	Bal	81
718	7.0	2.5	1.0	3.2	Bal	83, 84
719	7.0	2.5	1.0	0.8	Bal	85, 86
720	8.4	2.5	1.0	0.8	Bal	87, 88
721	8.4	2.5	1.0	1.6	Bal	89

TABLE 4. POWDER CONVERSIONS IN FOURTH QUARTER

XSR-Run No.	Alloy	Nozzle Dia in. (cm)	Cup Speed (K)	Cup Radius in. (cm)	Nozzle Temperature °F (°C)	Melt Temperature °F (°C)	Alloy Melting Point °F (°C)	Yield 140 Mesh (%)
78	716	0.100 (0.254)	24	3.125 (7.938)	1350 (732)	1475 (802)	1175 (635)	67.0
79	716	0.100 (0.254)	24	3.125 (7.938)	1300 (704)	1500 (816)	1160 (627)	74
80	716	0.100 (0.254)	24	3.125 (7.938)	1300 (704)	1500 (816)	1160 (627)	44
81	717	0.100 (0.254)	24	3.125 (7.938)	1300 (704)	1500 (816)	1190 (643)	49
83	718	(0.100) (0.254)	24	3.125 (7.938)	1300 (704)	1500 (816)	1185 (641)	74
84	718	0.100 (0.254)	24	3.125 (7.938)	1300 (704)	1500 (816)	1185 (641)	78
85	719	0.100 (0.254)	24	3.125 (7.938)	1300 (704)	1500 (816)	1185 (641)	70
86	719	0.100 (0.254)	24	3.125 (7.938)	1300 (704)	1500 (816)	1185 (641)	58
87	720	0.100 (0.254)	24	3.125 (7.938)	1380 (749)	1500 (816)	1180 (638)	70
88	720	0.100 (0.254)	24	3.125 (7.938)	1380 (749)	1500 (816)	1160 (627)	50
89	721	0.100 (0.254)	35	3.125 (7.938)	1380 (749)	1550 (843)	1160 (627)	81

The aluminum powder was screened in helium through −80 mesh and −140 mesh screens. The −140 mesh fraction was then outgassed, and 3-in. dia (7.6 cm) extrusion cans filled and sealed under vacuum. The cans were fabricated from 6061 Al alloy.

Nine aluminum alloy powder billets were direct-extruded at AFML during this report period. Six different Co- and Zn-modified 7075 compositions were extruded at 700°F and a 12:1 reduction ratio. Three billets were extruded at 650°F and the same reduction ratio to further evaluate the effect of extrusion temperature on tensile properties. The nominal and actual compositions of the extruded alloys is given in Table 5. It appears that some Zn and Co was lost during the melting and the atomization cycle for alloys containing high concentrations of these elements. However, the deviations from the nominal composition, as determined by atomic absorption spectrometry, were less than one wt % in all cases.

Table 6 gives details of all extrusions and shows them to be visually sound and without evidence of cracking or other metalworking defects. Final diameter was a nominal 0.84 in. (2.1 cm).

TABLE 5.   COMPOSITION OF ALLOYS EXTRUDED

XSR Run No.	VM No.	Analysis	Zn	Mg	Cu	Co	Al
43 and 44	630	Nominal	9.8	2.5	1.0	0.8	Bal
		Actual	8.8	2.2	1.0	0.8	Bal
75 and 76	678	Nominal	5.6	2.5	1.0	0.8	Bal
		Actual	5.2	2.55	1.05	0.8	Bal
78 and 79	716	Nominal	5.6	2.5	1.0	1.6	Bal
		Actual	4.9	2.2	0.9	1.0	Bal
81	717	Nominal	5.6	2.5	1.0	3.2	Bal
		Actual	5.4	2.45	1.0	2.25	Bal
83 and 84	718	Nominal	7.0	2.5	1.0	3.2	Bal
		Actual	6.4	2.5	1.0	2.5	Bal
85	719	Nominal	7.0	2.5	1.0	0.8	Bal
		Actual	6.6	2.4	1.0	0.8	Bal
87	720	Nominal	8.4	2.5	1.0	0.8	Bal
		Actual	7.4	2.4	1.0	0.8	Bal

TABLE 6.   ALUMINUM ALLOY EXTRUSION PARAMETERS

XSR Run No.	VM No.	Temperature °F	(°C)	Reduction Ratio	Breakthrough Pressure (ksi)	(MPa)	Maximum Running Pressure (ksi)	(MPa)	Visual Appearance
43-44	630	650	(343)	12:1	84	579	70	483	good
75-76	678	700	(371)	12:1	70	483	68	469	good
78	716	650	(343)	12:1	70	483	73	503	good
79	716	700	(371)	12:1	97	669	97	669	good
81	717	700	(371)	12:1	103	710	95	655	good
83	718	650	(343)	12:1	78	538	73	503	good
84	718	700	(371)	12:1	76	524	70	483	good
85	719	700	(371)	12:1	68	469	62	428	good
87	720	700	(371)	12:1	88	607	89	614	good

SECTION IV

MATERIAL EVALUATION

IRON ALLOYS

Metallographic examination of the iron alloy extrusions made last quarter was conducted for both the as-extruded and heat treated conditions. The austenitizing heat treatment was 2040°F (1115°C) for 10 min with an air cool followed by a 1010°F (543°C)/2 hr/AC tempering cycle. This is the standard heat treatment used for M50 bearings. This heat treatment was selected because of similarities between the M50 bearing steel and the XSR alloys produced.

Figure 1 shows the microstructures of an M50 bearing ball and inner race. The light etching phase is $(MoV)_6C$ carbides which are not uniformly dispersed due to casting segregation present in the starting VIM-VAR ingot. Subsequent hot forging and austenitizing heat treatments were not successful in dissolving these coarse, segregated carbides. The small, finely dispersed carbides may be $M_{23}C_6$ chromium-rich carbides which were fully dissolved during the austenitizing heat treatment and reprecipitated on cooling to ambient temperature. The matrix is a tempered martensite containing an undertermined amount of retained austenite.

Figure 2 shows the microstructure of M50 barstock made by direct extrusion of RSR powder, XSR Run No. 62. This structure is very fine and cannot be clearly resolved at a 1000X magnification. At a magnification of 10,000X, a uniform dispersion of carbides (0.3 to 0.6μm) was observed embedded in a tempered martensite matrix. The hardness of this microstructure was R_c62 compared to R_c61 for the conventional material. A duplex grain size ranging from 3 to 30μm was observed, indicating that the austenitization temperature of 2040°F (1115°C) could be lowered to produce a finer, more uniform grain size. The grain size of the as-extruded material was not determined.

The carbide dispersion in 140 mesh powder particles of these alloys was not resolved using scanning electron microscopy techniques. Occasional powder particles contained clusters of coarse carbides, but it is believed that these were present in the melt prior to atomization. Extrusion of the powder, i.e., XSR 62 (M50) at 15:1 and 1650°F resulted in uniform precipitation of carbides, Figure 3. These carbides were not changed appreciably by austenitization and tempering heat-treat cycles.

Two extrusion temperatures, 1650°F (899°C) and 1750°F (954°C), were evaluated for their effect on the final microstructure of M50 and modified M50 steel. There was virtually no difference in alloy microstructure or hardness as a result of the variation in extrusion temperature, Table 7. The 1650°F (899°C) extrusion temperature was the lowest possible at a 12:1 reduction ratio because of extrusion press limitations.

A reduction in the Mo concentration from 4.2 to 2.0 wt % in M50 (VM 679) greatly reduces the number of Mo_6C carbides present in the tempered martensite matrix. The quenched and tempered hardness was also lowered from R_c 62.5 to R_c 59. The carbide size (0.2 to 0.6μm) and the martensitic matrix remained the same. A 5.0 wt % Cr addition was made to the VM 679 alloy (VM 680) while the carbon level was increased from 0.80C to 0.95C to increase the volume fraction of carbides. Figure 4 shows a large number of (0.4 to 1.0μm) $M_{23}C_6$ Cr-rich carbides in the grain boundaries of the quenched and tempered martensite. The smaller, less prevalent matrix carbides are $(MoV)C_6$. The Cr and C addition raised the hardness to R_c 61.5 from R_c 59. The grain size of this alloy was a fine 3 to 6μm.

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FIGURE 1
VIM/VAR M50 BEARING STEEL
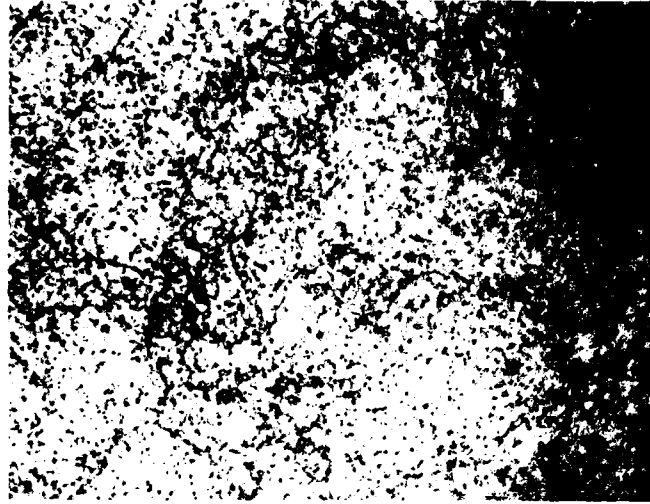### (A) Inner Race (B) Ball Bearing

Mag: 1000X
Etch: Glyceregia

A

Mag: 1000X
Etch: Glyceregia

B

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FIGURE 2
XSR 62 (M50 BEARING STEEL)
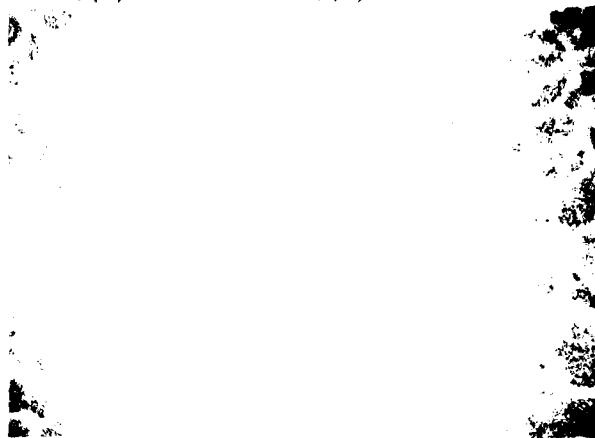

Mag: 1000X
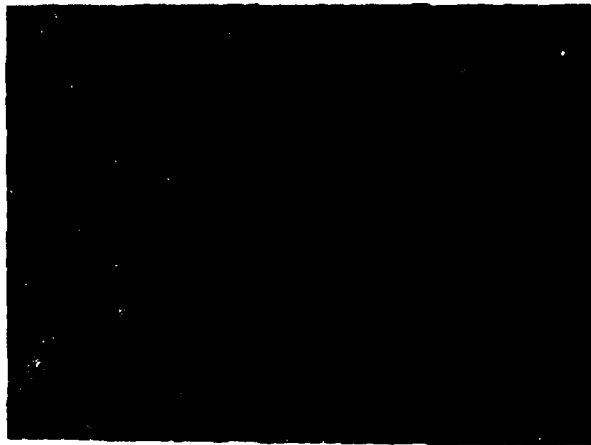Etch: Glyceregia

A

Mag: 10,000X
Etch: Glyceregia

B

FD 175211

8

FIGURE 3
MICROSTRUCTURAL CHANGES IN XSR 62 (M50) BEARING ALLOY DURING PROCESSING:
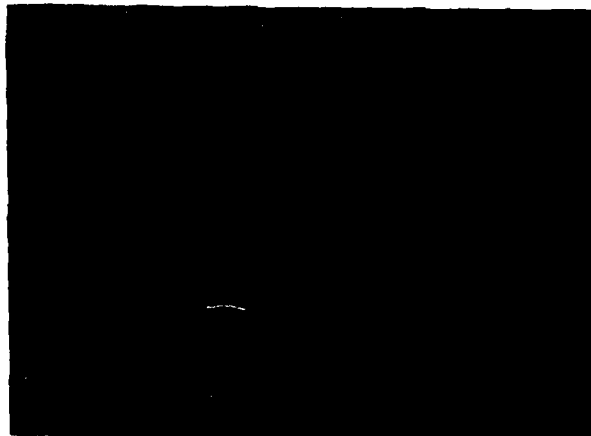(A) 95 μm dia Powder Particle, (B) Direct Extrusion, (C) Heat Treated After Extrusion

Mag: 6000X
Etch: Picral

A

Mag: 10,000X
Etch: Picral

B

Mag: 10,000X
Etch: Picral

C

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TABLE 7.   HARDNESS OF IRON ALLOY 12:1 DIRECT EXTRUSIONS, R_c

Sample ID XSR	VM No.	Extrusion Temperature °F (°C)	Composition (Nominal)	As-Extruded	2040°F (1115°C) 10 min/AC + 1010°F (543°C) 2 hr/AC
62	681	1650 (899)	Fe-0.8C-4Cr-4.2Mo-1V-0.2Mn-0.15Si	58	62
63	681	1750 (954)	Fe-0.8C-4Cr-4.2Mo-1V-0.2Mn-0.15Si	60	63
53	649	1750 (954)	Fe-0.8C-4Cr-4.2Mo-1V-0.45Mn-0.3Si	61	63
61	679	1650 (899)	Fe-0.8C-4Cr-2Mo-1V-0.2Mn-0.15Si	57	59
58	679	1750 (954)	Fe-0.8C-4Cr-2Mo-1V-0.2Mn-0.15Si	56	59
60	680	1650 (899)	Fe-0.95C-9Cr-2Mo-1V-0.2Mn-0.15Si	55	61
59	680	1750 (954)	Fe-0.95C-9Cr-2Mo-1V-0.2Mn-0.15Si	58	62
52	650	1650 (899)	Fe-1.15C-14.75Cr-4Mo-1.2V-0.2Mn-0.15Si	54	60
54	650	1750	Fe-1.15C-14.75Cr-4Mo-1.2V-0.2Mn-0.15Si	56	61
Conventional M50			Fe-0.8C-4Cr-4.2Mo-1V-0.2Mn-0.1Si		61

FIGURE 4
COMPARISON OF XSR M50 AND MODIFIED BEARING ALLOYS
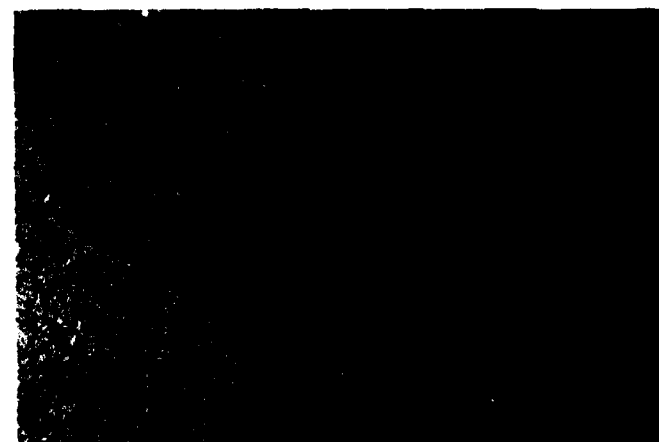(A) XSR 62 (M50), (B) XSR 58 (M50-2.2 Mo), (C) XSR 60 (XSR 58 + 5% Cr)

Mag: 3000X
Etch: Nital

A

Mag: 3000X
Etch: Nital

B

Mag: 3000X
Etch: Nital

C

11

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Finally, an alloy (VM 650) with the composition (Fe-1.15C-14.75Cr-4.0Mo-1.2V-0.2Mn-0.15Si) was evaluated. Figure 5 shows that this alloy contains an extremely large volume fraction of carbides. Most of the carbides are present in the grain boundaries in the form of 1 to 2μm M_6C particles. It appears that there are fewer M_6C carbides present than in the other alloy modifications of M50. The nominal hardness of this alloy was R_c 60.5 and the grain size ranged from 1.5 to 3μm.

All of the M50 and modified alloys showed a uniform fine dispersion of carbides which was maintained from the extruded to the quenched and tempered barstock. The rolling contact fatigue and wear resistance of these alloys should be superior to their conventionally-processed counterparts providing foreign inclusions are minimized. All of the M50 modified alloys contained an unacceptable number of ceramic inclusions due to the firability of some of the ceramics employed in the atomizing system. The system has since been redesigned to reduce ceramic content to acceptable levels. Roller bearings for rolling contact fatigue tests will be manufactured when the steels meet the requirements of the Jernkontoret Inclusion Chart, as specified in ASTM E45.

ALUMINUM ALLOYS

Evaluation of the Co- and Zn-modified 7075 Al alloy matrix was completed during this quarter. As reported last quarter, a coarse primary dispersion of Co_2Al_9 was present in all powders and in subsequent extrusions due to nucleation of this phase in the liquid metal prior to atomization. X-ray image microscopy revealed the cobalt in the extrusions, Figure 6.

The above results were obtained with a rotary atomization system which required a critical heat balance for stable operation and optimum powder yield (normally 50 to 70°). Since the results were less than satisfactory, modifications were required which would permit the use of the *higher melt temperatures required for complete solutioning of the cobalt*. With a modified atomizer and other system changes, melt temperatures were increased 200 to 300°F (to 1850°F or 982°C), and powder yields to over 70° of -140 mesh powder were achieved. The Co-modified 7075 alloy produced contained the very fine Co_2Al_9 dispersion which is shown in Figure 7. With this system, Al alloys containing high concentrations of transition elements and other elements soluble only at very high liquid metal temperatures will be possible. The systematic examination of 7075 or 7178 alloys with iron additions will be completed during the next reporting period. It is expected that routine X-ray diffraction lattice parameter measurements on powders will allow element supersaturation to be determined quantitatively in the high-temperature binary alloys previously listed in Table 2.

Despite the Co_2Al_9 segregation initially obtained, hardness, tensile, and stress-rupture tests were performed on these alloys. Hardness varied from 87 to 95R_B in the T6 heat-treated condition and 82 to 91R_B in the T7 heat-treated condition. There did not appear to be any correlation between hardness and extrusion temperature or composition, as shown in Table 8.

Room temperature tensile properties for many alloys were excellent when compared to Al 7075T6, Table 9. The improved tensile properties were a result of increased Zn additions above the 5.6 wt % level present in 7075. The 0.2% yield strength of a 0.8 wt % Co-modified alloy was raised to 84.5 ksi (583 Mn/m²) at 5.6 wt % Zn and to 105.5 ksi (727 Mn/m²) at 9.8 wt % Zn. This strength increase was due solely to an increase in the amount of $MgZn_2$ and other complex age-hardening precipitates. Overaging these alloys by employing a 330°F (160°C)/8 hr/AC heat treatment (T7) drastically reduced the 0.2% yield strength of all the alloys. Only a slight difference in strength is apparent between the 5.6 wt % Zn and 9.8 wt % Zn alloy, Figure 8. As was expected, varying Co concentrations had virtually no effect on tensile properties since most of the Co was in coarse precipitates. These results indicate that the good room temperature tensile properties of high Zn- and Co-modified 7075 alloys are due to high Zn concentrations rather than to Co additions.

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FIGURE 5
VM 650 ALLOY (Fe-1.15C-14.75Cr-4Mo-1.2V-0.2Mn-0.15Si)

Mag: 3000X
Etch: Nital

A
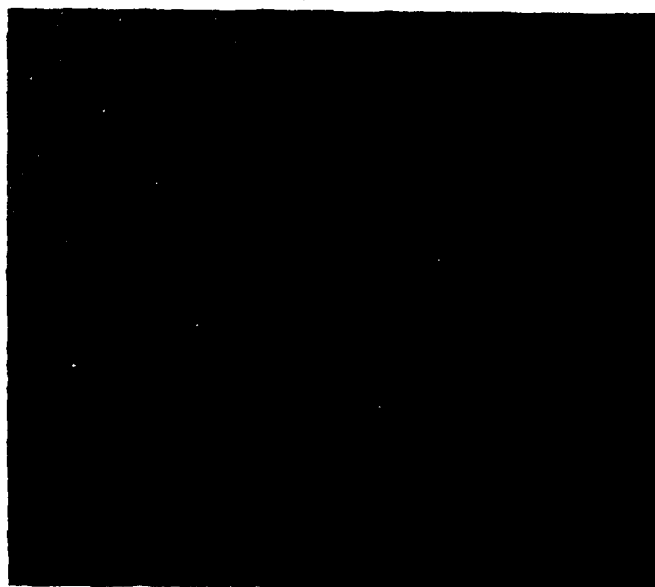
Mag: 20,000X
Etch: Nital

B   FD 175214

FIGURE 6
CO_2AL_9 SEGREGATION IN CO MODIFIED AL 7075 ALLOY EXTRUSION
(A) Scanning Electron Micrograph (B) X-Ray Image
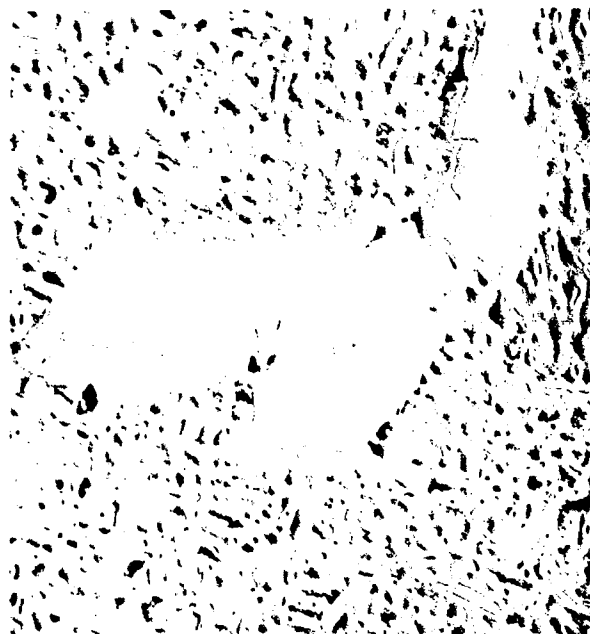
Mag: 250X

A

Mag: 250X

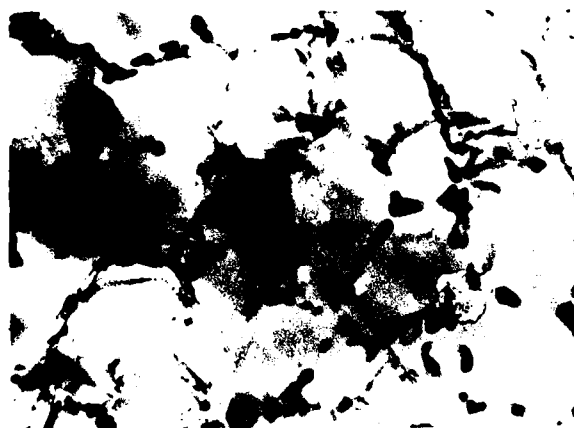B

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14

FIGURE 7
DARK ROD-LIKE PHASE IS CO_2AL_9 WHICH SEGREGATED AT GRAIN BOUNDARIES IN AL-8.4ZN-2.5MG-1.0CU-O.8 CO ALLOY AFTER HOT COMPACTION AT 700°F (371°C)
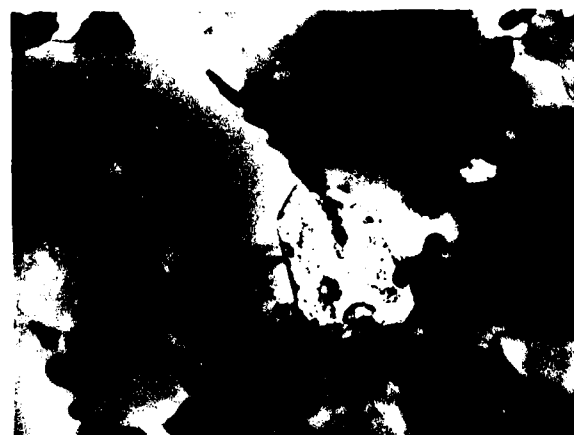

Mag: 13,500X

A

Mag: 22,000X

B

Mag: 37,000X

C

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TABLE 8. HARDNESS OF ALUMINUM ALLOY EXTRUSIONS, R_B

ID XSR	VM No.	As-Extruded	870°F (466°C)/1 hr/WQ +250°F (121°C)/26 hr/AC T6	870°F (466°C)/1 hr/WQ +250°F (121°C)/26 hr/AC + 330°F (166°C)/8 hr/AC T7
43-44	630	53	93	86
75-76	678	47	93	88
78	716	56	87	82
79	716	49	89	85
81	717	51	90	85
83	718	60	95	91
84	718	50	93	88
85	719	54	92	86
87	720	49	93	87

TABLE 9. ROOM TEMPERATURE TENSILE PROPERTIES OF ALUMINUM ALLOYS IN THE T6 AND T7 HEAT-TREATED CONDITIONS*

ID XSR	VM No.	Extrusion Temperature °F (°C)	Heat Treatment	0.2% Yield Strength (ksi)	Tensile Strength (ksi)	Elongation (%)	RA (%)
43-44	630	650 (343)	T6	102.8	106.5	6.7	10.0
43-44	630	650 (343)	T7	77.1	84.7	9.0	20.0
75-76	678	700 (371)	T6	84.8	91.2	7.6	15.8
75-76	678	700 (371)	T7	75.8	82.2	8.0	21.7
78	716	650 (343)	T6	80.8	87.7	8.0	14.4
78	716	650 (343)	T7	73.5	79.8	7.6	18.8
79	716	700 (371)	T6	82.4	87.8	10.7	25.8
79	716	700 (371)	T7	75.2	82.0	11.0	37.2
81	717	700 (371)	T6	90.1	96.4	5.3	11.3
81	717	700 (371)	T7	77.1	83.3	7.0	15.0
83	718	650 (343)	T6	101.2	104.4	3.3	8.8
83	718	650 (343)	T7	82.7	88.2	4.0	12.5
84	718	700 (371)	T6	97.5	101.1	4.0	8.3
84	718	700 (371)	T7	80.5	86.1	6.0	12.2
85	719	700 (371)	T6	95.9	102.2	8.0	15.5
85	719	700 (371)	T7	77.2	83.7	10.0	29.1
87	720	700 (371)	T6	100.1	104.7	8.4	18.0
87	720	700 (371)	T7	81.0	85.7	10.0	33.5

*T6   870°F (466°C)/1 hr/WQ + 250°F (121°C)/26 hr/AC
T7   870°F (466°C) 1 hr/WQ + 250°F (121°C)/26 hr/AC + 330°F (166°C)/8 hr/AC
Data represents average of two tests.

FIGURE 8
ROOM TEMPERATURE 0.2% YIELD STRENGTH vs ZINC CONCENTRATION FOR
0.8 WT% COBALT MODIFIED 7075 ALLOYS IN THE T6 AND T7
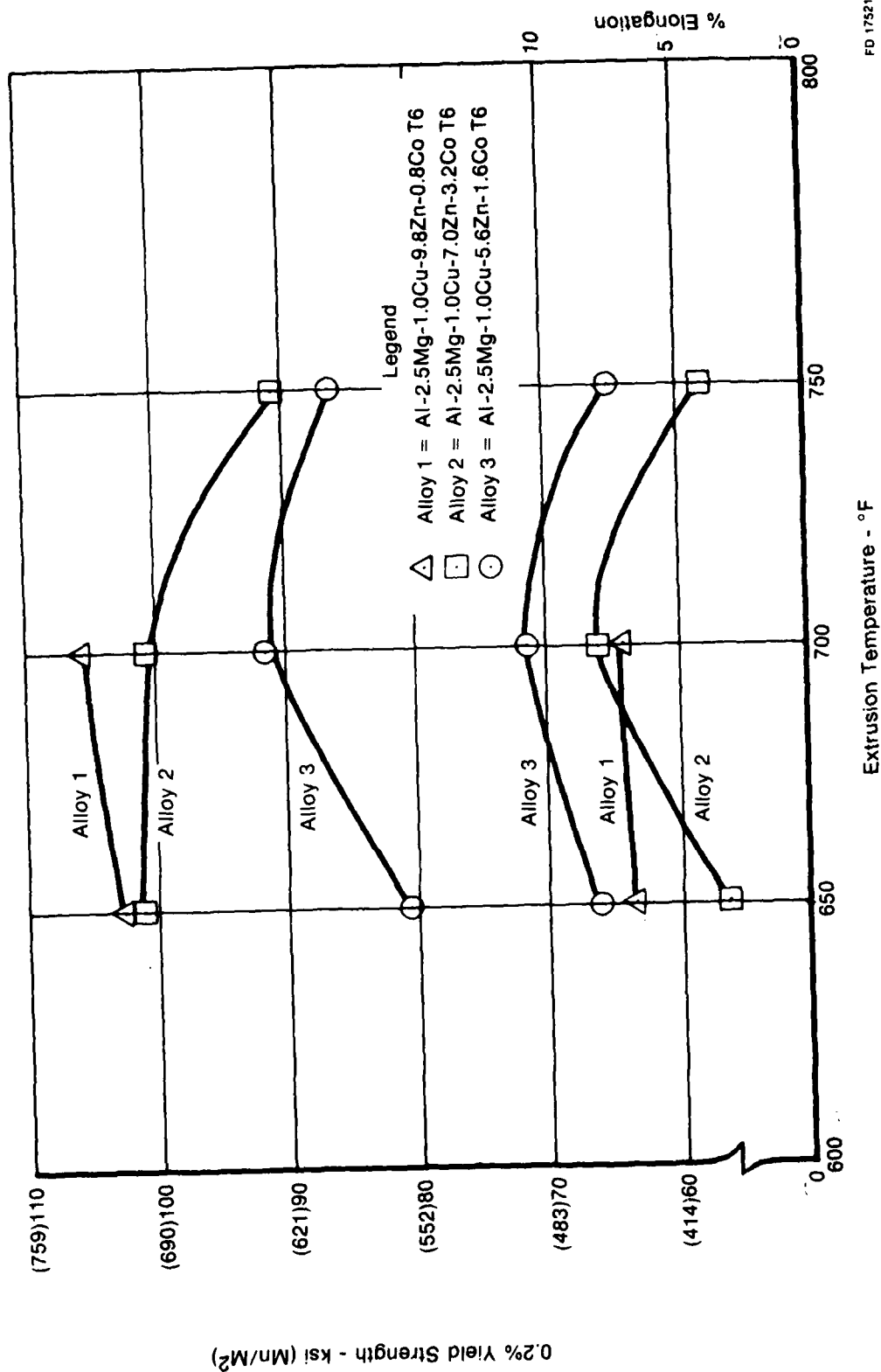## HEAT TREATED CONDITIONS
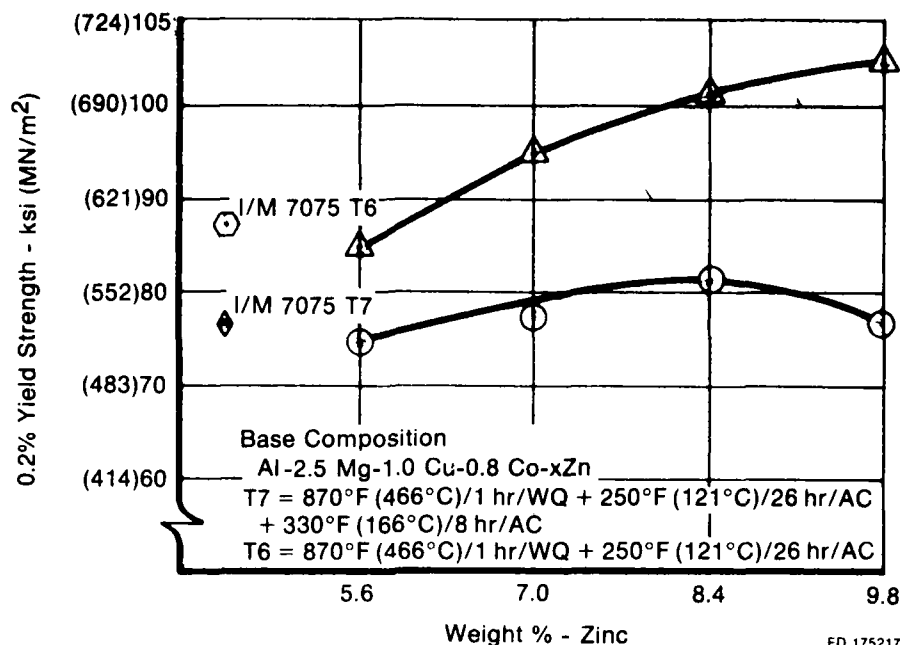
Four stress-rupture tests were performed on two alloys with a high Co concentration (2.5 wt %) XSR 81 (VM 717) and XSR 83 (VM 718). The test conditions were 300°F — 30 ksi and represented both the T6 and T7 heat treatments. Stress-rupture lives ranged from 51.9 hr for XSR 83T7 to 74.5 hr for XSR 83T6, Table 10. These values are typical of the 7075T6 alloy without the addition of Co.

As previously reported, a reduction in Al alloy extrusion temperature from 800°F (427°C) to 700°F (371°C) for a given composition improved the tensile properties. Further reducing the extrusion temperature to 650°F (343°C) resulted in no additional improvements in yield strength and significantly reduced tensile ductility in some alloys, Figure 9.

TABLE 10. STRESS-RUPTURE PROPERTIES OF TWO ALUMINUM ALLOYS

ID XSR	VM No.	Heat Treatment	Temperature °F (°C)	Stress ksi (MN/m²)	Rupture Life (hr)	Elongation (%)
81	717	T6	300 (149)	30 (207)	56.0	16.3
81	717	T7	300 (149)	30 (207)	60.8	18.3
83	718	T6	300 (149)	30 (207)	74.5	16.4
83	718	T7	300 (149)	30 (207)	51.9	15.0

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FIGURE 9
ROOM TEMPERATURE 0.2% YIELD STRENGTH AND % ELONGATION vs EXTRUSION TEMPERATURE
FOR COBALT AND ZINC MODIFIED 7075 T6 ALLOYS

Legend

Alloy 1 = Al-2.5Mg-1.0Cu-9.8Zn-0.8Co T6
Alloy 2 = Al-2.5Mg-1.0Cu-7.0Zn-3.2Co T6
Alloy 3 = Al-2.5Mg-1.0Cu-5.6Zn-1.6Co T6

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A study was made on Al alloy (Al-2.5Mg-1.0Cu-7.0Zn-0.8Co) to determine the effect of inert vs air powder handling on resultant tensile properties. Alloy XSR 30-1 was sieved and stored in air after atomization while alloy XSR 85 was sieved and stored in vacuum. Both alloys were extruded at 700°F (371°C) and given the same T6 heat treatment. The vacuum-processed material showed an average 3.7 ksi (25.5/Mn/m²) yield strength and a 5.8 ksi (40Mn/m²) tensile strength advantage. However, tensile elongation dropped slightly from 9.3 to 8.0%, Table 11. Since the microstructures of the two XSR powders were similar, the difference in properties may be due to a difference in alloy chemistry.

The stress corrosion cracking resistance of these alloys was not investigated due to their nonrepresentative nature. Apparatus for measuring stress corrosion cracking resistance has been procured, and these tests will be performed on future alloy compositions.

TABLE 11. ROOM TEMPERATURE TENSILE PROPERTIES OF AIR VS VACUUM-HANDLED XSR POWDER

ID XSR	VM No.	Powder Handling	Heat Treatment	0.2% YS (ksi)	Tensile Strength (ksi)	Elongation (%)
30-1	595	Air	T6	92.3	96.4	9.3
30-1	595	Air	T6	92.3	96.4	9.3
30-1	595	Air	T6	92.1	96.4	9.3
85	719	Vacuum	T6	95.3	100.0	8.0
85	719	Vacuum	T6	96.4	104.4	8.0

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SECTION V

ONGOING STUDY

In the next quarter, powder conversions will be made for the second Al alloy matrix. Billets containing Co- and Fe-modified Al 7075/7178 alloy powder with a fine intermetallic dispersion will be extruded and evaluated. Bearing alloys from the first matrix will be atomized using an improved atomizing system.